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Superplastic behavior of coarse-grained aluminum alloys

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Abstract

In this paper we concentrate on the superplastic behavior and the microstructural evolution of two coarse-grained Al alloys: Al–4.4w/oMg and Al–4.4w/oMg–0.4w/oCu. The values for the strain rate sensitivity index and activation energy suggest that solute drag on dislocation motion is an important phenomenon. The plasticity of these materials is further enhanced by a reconstruction mechanism due to dynamic recovery and recrystallization. At temperatures lower than 450 °C and for strain rates up to 10^{-2} s^{-1} the main reconstruction mechanism can be described by dynamic recovery. The dynamic recovery compensates the strain hardening and the deformation takes place in a steady state like mode resulting in high values for the maximum tensile elongation (260% for Al–Mg and 320% for Al–Mg–Cu). The deformed microstructure is characterized by a lower value of the average grain size and an increased density of low angle grain boundaries as compared to the original materials. At higher strain rates dynamic recrystallization produces oscillations of the flow stress curves or a coarser microstructure.

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Keywords: Superplasticity; Coarse-grained aluminum alloys; Dynamic recovery and recrystallization texture

1. Introduction

Coarse-grained superplastic alloys respond in a unique way to industrial demands by showing high elongation to failure (>200%) at high strain rates ($>10^{-2} \text{ s}^{-1}$) combined with reduced costs of the materials [1–3]. Such alloys can lead to the transition from niche applications, where currently “conventional” superplastic forming of fine-grained alloys is used, towards volume component production [4].

The main mechanism responsible for superplasticity in coarse-grained Al–Mg based alloys is the solute drag creep [5]. In this case the strain rate sensitivity index m , which dictates the necking instability, has a value of 0.3. This value is lower than in the case of the fine-grained materials, where $m = 0.5$ [6]. As a consequence the coarse-grained materials will show in general lower values of the maximum tensile elongation. Nevertheless, the main advantage of superplasticity based on solute drag creep is that this mechanism has virtually no grain size dependence and therefore the preparation of such materials is less complex.

Judging the superplastic behavior of coarse-grained materials only in terms of solute drag creep is actually an over-

simplification. This is because plastic deformation at a high temperature is accompanied by dynamic recovery and recrystallization. The microstructure undergoes important modifications leading to different microstructures that show less and less resemblance to the original. These processes, that are grain size-dependent, might influence the values of the maximum tensile elongation that can be obtained in a coarse-grained superplastic alloy.

In the present study we focus on an examination of the mechanical behavior and microstructural evolution in two coarse-grained Al alloys. First the parameters characterizing the superplastic deformation are determined, i.e. the strain rate sensitivity index and the activation energy for plastic flow. Next, the changes in the microstructure are revealed using orientation imaging microscopy (OIM).

2. Experimental

In this study two materials have been used: an Al–4.4Mg alloy and an Al–4.4Mg–0.4Cu alloy (in w/o). Specimens for tensile testing were laser cut from 2 mm thick cold-rolled metal sheets with the gauge direction parallel to the rolling direction and subsequently annealed for 10 min at 450 °C before deformation. Tensile elongation measurements were performed at constant

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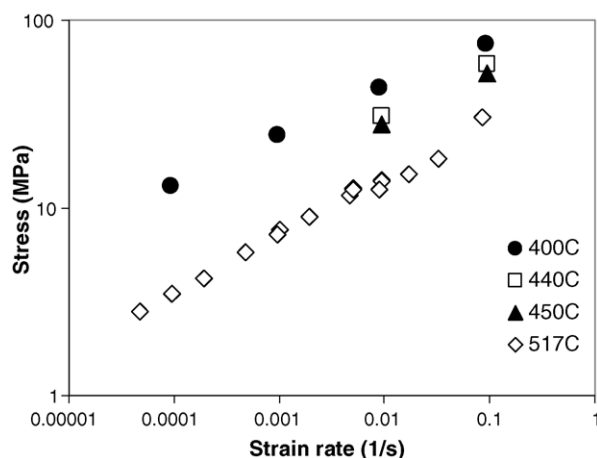


Fig. 1. The temperature dependence of the flow stress vs. strain rate in Al-Mg-Cu.

crosshead speed under controlled temperature conditions using an Instron tensile machine and a three-zone-split furnace. The strain rate sensitivity index, m , and the activation energy, Q , were determined from strain-rate-change (SRC) tests. Two types of specimens were used for microstructural investigations. Several regions of a specimen deformed to fracture (325%) were investigated for large values of the strain (>1.5) and regions of partially deformed and subsequently quenched specimens were used for low strain values (<1.5). The specimens deformed to fracture were removed from the three-zone-split furnace after approximately 10 min. The partially deformed samples were quenched using a pressurized airflow with a cooling rate of the order of $25^\circ/\text{s}$ for the temperature interval between 450 and 250°C and with about $2^\circ/\text{s}$ from 250°C to room temperature (RT). Subsequent annealing experiments (not shown) indicate that the microstructure formed during high temperature deformation is stable during annealing at temperatures as high as 450°C . Consequently the relative low cooling rates used for the samples in this study have little effect on altering the microstructure of the investigated specimens. The strain value at the position where the microstructure was investigated was determined from optical measurements of the cross-sectional area of the gauge of the deformed specimen. The grain structure of the material was determined using the orientation imaging microscopy (OIM) facility attached to a scanning electron microscope (SEM), Philips XL30-S FEG.

3. Results and discussion

The maximum values of the elongation to failure (260% for the Al-Mg alloy and, respectively, 320% for the Al-Mg-Cu alloy) were obtained for tensile tests performed at 450°C and at a strain rate of 10^{-2} s^{-1} [7,8]. Other combinations of temperature and strain rates resulted always in lower values for the elongation to failure. In the case of Al-Mg-Cu alloy the variation of the flow stress as a function of the initial strain rate for several temperatures is presented in Fig. 1. Here the values of the true stress are taken at a strain of 0.1. For the temperatures used in this study the stress and strain rate have a power law

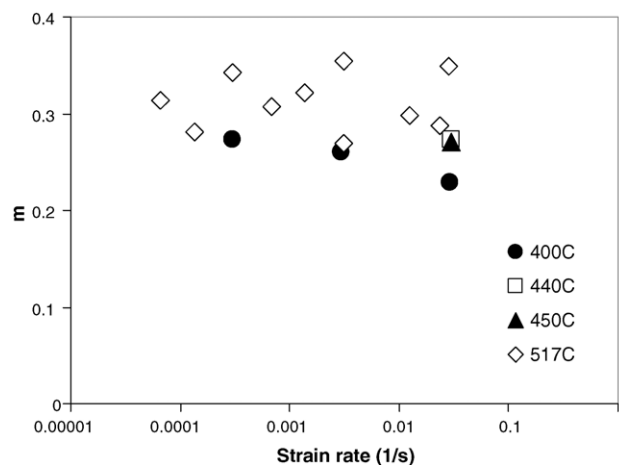


Fig. 2. The dependence of the strain rate sensitivity index (m) on temperature and strain rate for Al-Mg-Cu.

relationship, i.e. $\sigma \propto \dot{\epsilon}^m$, where m is the strain rate sensitivity index. This relationship is valid for the entire temperature and strain rate intervals investigated (from 400 to 517°C and from 10^{-4} to 0.1 s^{-1} , respectively). The values of m as a function of strain rate and temperature are presented in Fig. 2. At 517°C the m values are scattered around 0.3 for the considered strain interval while at lower deformation temperature (400°C) the strain rate sensitivity tends to decrease with increasing strain rate. The activation energy determined from the data from Fig. 1 is equal to 153 kJ/mol at a stress value of 25 MPa . This value is close to the activation energy for diffusion of Mg in Al (136 kJ/mol) [5]. Similar results were obtained for the Al-Mg alloy. The values of the strain rate sensitivity index and that of the activation energy indicate that solute drag on dislocation gliding is rate-controlling mechanism in these alloys.

A set of representative flow stress curves for the Al-Mg-Cu alloy deformed in uniaxial tension at 400°C are presented in Fig. 3. In all three cases the flow stress increases to a peak and then decreases sharply over a stress interval of about 0.05%. The high initial strength can be correlated with an initial low density of mobile dislocations and the flow stress decrease with dislocation multiplication and cross-slip [9]. After this peak the flow stress reaches a steady state value and a slight decrease

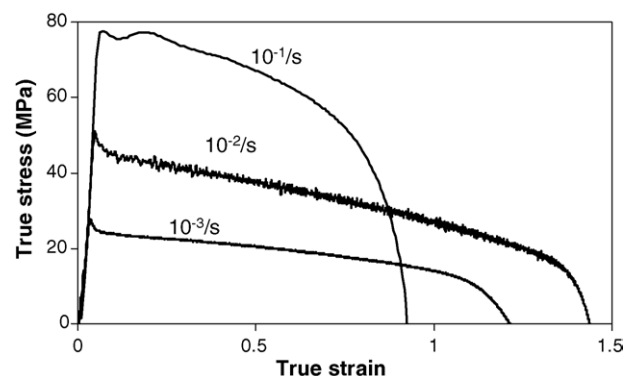


Fig. 3. Representative flow stress curves in uniaxial tension for the Al-Mg-Cu alloy deformed at 400°C at three different strain rates.

of the stress with increasing strain is observed in Fig. 3. The tests were performed at constant crosshead speed. The steady state, which is similar to creep experiments, indicates the existence of restoration mechanisms like dynamic recovery and/or dynamic recrystallization (DRX). The additional oscillation of the flow stress observed in the case of the deformation performed at 10^{-1} s^{-1} might be related to the occurrence of dynamic recrystallization [10]. At lower strain rates (10^{-2} or 10^{-3} s^{-1}) the flow stress curves suggest that the dynamic recovery is the main restoration mechanism [9]. This suggestion is supported further by the decrease in the average grain size accompanied by the increase of the fraction of low angle grain boundaries as the plastic strain increases (Fig. 4). The dislocations generated during plastic deformation partly annihilate and partly rearrange in sub-boundaries. With further deformation more dislocations are absorbed by the sub-boundaries leading to an increase in the misorientation between the subgrains and the transformation of some of the sub-boundaries into grain boundaries. The newly

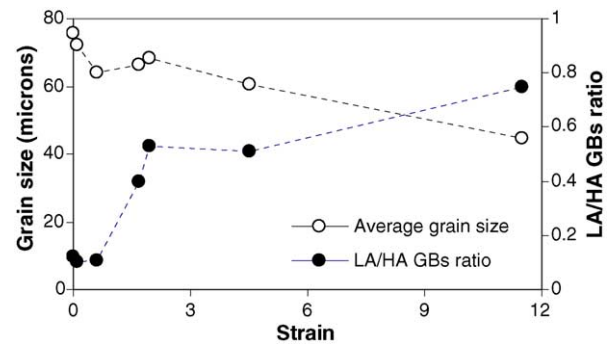


Fig. 4. Evolution of average grain size and of the ratio of low angle grain boundaries/high angle grain boundaries as a function of plastic strain.

generated microstructure is characterized by a finer grain size and an increased grain boundary density as compared to the original material. A typical example of the deformed microstructure can be seen to the right side of Fig. 5.

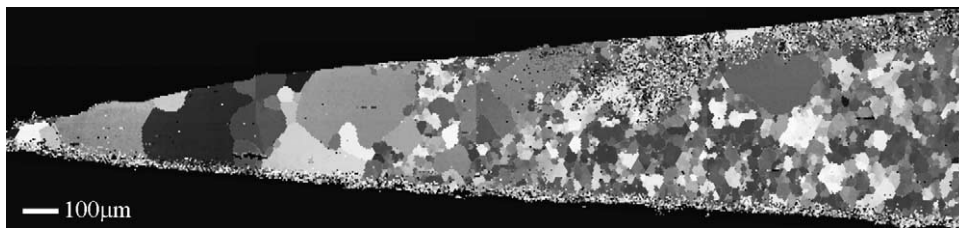


Fig. 5. Example of the microstructure close to the fracture place, in a plane parallel to the deformation direction and perpendicular to the plane of the metal sheet, revealed by OIM for the Al-Mg-Cu alloy deformed at 440°C and at an initial strain rate of 10^{-2} s^{-1} . On the right side can be seen the typical fine microstructure formed during superplastic deformation assisted by dynamic recovery. On the left side, close to the fracture tip, coarse grains were formed by dynamic recrystallization and strain induced grain growth.

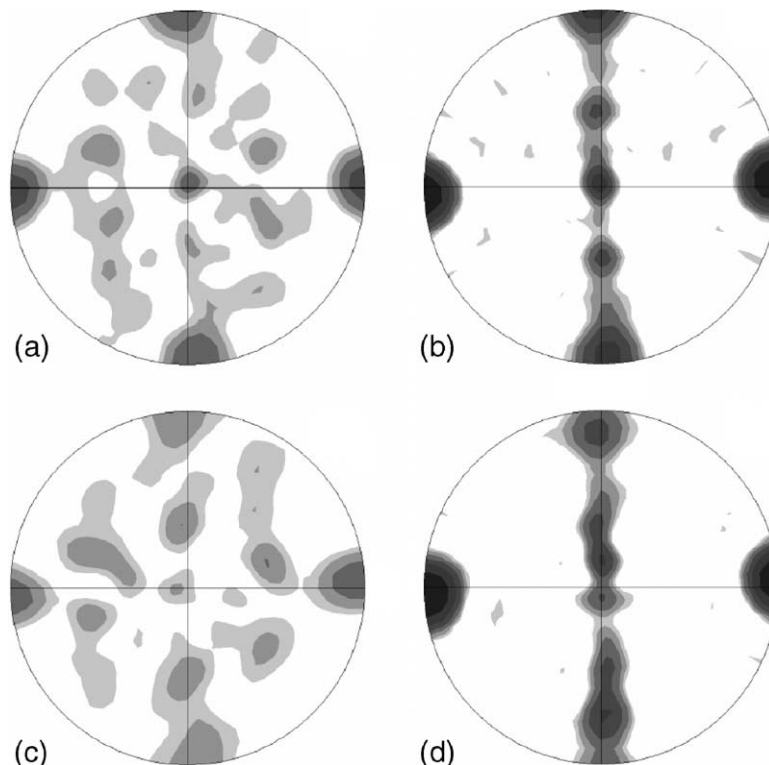


Fig. 6. Pole figures of the $\langle 001 \rangle$ orientation for Al-Mg (a) before deformation and (b) deformed 60% at 440°C and 10^{-2} s^{-1} ; and for Al-Mg-Cu (c) before deformation and (d) deformed 60% at 440°C and 10^{-2} s^{-1} . The deformation direction was parallel to the horizontal axis.

The coarse-grained Al–Mg and Al–Mg–Cu alloys fail by prolonged necking and ductile fracture. Close to the fracture place the strain rate increases significantly leading to a local increase of the substructure density resulting in the appearance of dynamic recrystallization. This effect would account for the appearance of the coarse grains observed near the vicinity of the fracture tip (Fig. 5).

Another structural effect observed is the modification of texture. The starting material has a predominant cube texture (Fig. 6a and c). The materials deformed at a strain of 0.6 or higher present a strong $\langle 100 \rangle$ fiber texture with the fiber direction parallel to the deformation direction (Fig. 6b and d).

4. Conclusions

The principal mechanism of plasticity in the Al–Mg and Al–Mg–Cu alloys investigated in this study is affected by solute drag on gliding dislocations, as indicated by a strain rate sensitivity index of the order of 0.3 and a value of the activation energy close to that for the diffusion of Mg in Al. The plasticity of these materials is further enhanced by reconstruction mechanisms like dynamic recovery and dynamic recrystallization. At temperatures lower than 450 °C and for strain rates up to 10^{-2} s^{-1} the main restoration mechanism can be described as dynamic recovery. The dynamic recovery compensates the strain hardening and the deformation takes place in a steady state mode resulting in higher values for the maximum tensile elongation (260% for Al–Mg and 320% for Al–Mg–Cu). The

deformed microstructure is characterized by a lower value of the average grain size and an increased density of low angle grain boundaries, as compared to the original materials. At higher strain rates or close to the fracture place dynamic recrystallization produces oscillations in the flow stress curves and a coarser microstructure.

Acknowledgement

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References

- [1] M. Otsuka, S. Shibasaki, M. Kikuchi, *Mater. Sci. Forum* 233–234 (1997) 193–198.
- [2] W.J. Kim, *Mater. Sci. Forum* 304–306 (1999) 273–278.
- [3] H. Uchida, M. Asano, H. Yoshida, *Mater. Sci. Forum* 304–306 (1999) 309–314.
- [4] R.A. Friedman, S.G. Luckey, *Mater. Sci. Forum* 447–448 (2004) 199–204.
- [5] O.D. Sherby, E.M. Taleff, *Mater. Sci. Eng. A* 322 (2002) 89–99.
- [6] D.H. Bae, A.K. Ghosh, *Acta Mater.* 48 (2000) 1207–1224.
- [7] A.R. Chezan, J.Th.M. De Hosson, in: G. Bernhart, T. Cutard, P. Lours (Eds.), *Proceeding of the Third European Conference on Super Plastic Forming*, Ecoles des Mines d’Albi-Carmaux, Euro SPF 04, Cepadues-Editions, France, July 7–9, 2004, pp. 33–38.
- [8] A.R. Chezan, J.Th.M. De Hosson, *Mater. Sci. Forum* 495–497 (2005) 883–888.
- [9] G.A. Henshall, M.E. Kassner, H.J. McQueen, *Metall. Trans.* 23A (1992) 881–889.
- [10] J. Kroc, K. Balik, P. Lukac, *Mater. Sci. Eng. A* 234–236 (1997) 936–939.